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Notched Bar Testing.

As an outcome of the Congress of the International Association for Testing Materials held in London in April last, a Joint Committee on Materials and their Testing was formed under the chairmanship of Dr. Gough, with the co-operation of twenty-four technical institutions. One object of the Committee was to preserve interest in the work of the Association during the periods which elapse between the successive International Congresses.

The first technical discussion under the auspices of the newly formed Joint Committee has been arranged to take place in Manchester to-day, October 29th. The subject chosen, "Notched Bar Impact Testing," is one which has received a great deal of attention at intervals during the last twenty years, and has consequently been much discussed. It may be appropriate to recall some of the stages through which this discussion has passed.

The single-blow notched bar impact test was originally introduced largely on the initiative of armament manufacturers who were directly interested in the "shock-resisting" properties of steel. It was primarily regarded as a shock test, but its wider application led to a universal recognition (before 1920) that the essential feature of the test was the presence of the notch. The search then began for some means of comparing results obtained with different test pieces under different conditions of test, and continued intensively with the object of finding some method of comparison based on dimensional similarity, both of test pieces and machines. The simple application of the principle of geometrical similarity was shown, largely by the work of Stanton and Batson, to be inadmissible, and the analysis of the mechanism of failure of the notched bar test piece by Moser and others, from 1923 onwards showed clearly why this was so. At the meeting of the New International Association for Testing Materials at Zürich in 1931, the position disclosed was that in all countries (with the possible exception at that time of the United States of America) the value of the test for investigation and inspection of materials was being widely recognised. The notched bar impact test was being used to ensure that material accepted was in the best condition to effect a redistribution and relief of any localised stress (whether due to impact, notch effect, internal flaws, or inclusions) by local plastic deformation. To the metallurgist its value lay mainly in the fact that it provided a positive method of excluding material which was defective owing to overheating, inadequate annealing, temper brittleness, ageing embrittlement, and many other conditions which were known to be undesirable. But along with the growth of appreciation of the test all countries had been developing their own methods of carrying it out, and the time was evidently ripe for some attempt at international standardisation of the test piece.

Since that date, in spite of much work in all countries, progress towards international standardisation has been slow. The notched bar test does not yield a numerical value which can be utilised in design; but the impact figure, like elongation in the tensile test, should, for a given class of material, always exceed some arbitrary value chosen to give security in service. There is a general feeling, probably well founded, that each of the proposed test pieces is adequate for the purpose, and in all countries the disadvantages of lack of continuity of experience have to be weighed carefully against the possible advantages of a change. At the Manchester meeting, Dr. Max Moser will be able to put forward the Continental view of the position, though his contribution and the discussion generally will no doubt take a wider view of the subject. Elucidation of the mechanism and meaning of the test are of greater importance than standardisation.

The Oxford machine, from which fresh light on the test is confidently expected, seems to have proved a disappointment to some because its classification of materials is so very similar to that of the Izod machine in spite of the different types of stress imposed on the two test pieces. This only further emphasises the importance of the notch in localising the stress, however this may be directly applied to the test piece. It is at the moment when this localised stress exceeds the elastic limit of the material at the notch that the energy absorption, measured in the test, begins to take place. The important metallurgical problem is to account for the difference in behaviour, under localised stress, of materials which are sometimes apparently identical in other respects, and to explain why the notched bar impact value is often unique in indicating a condition, which, as a result of wide experience, is known to involve increased risk of premature failure of a material in service.

Progress in Metallurgical Microscopy.

By J. H. G. MONYPENNY, F. Inst. P.

IN metallography, as in other branches of science, which are concerned with a study of micro-structures progress is largely dependent on the capacity of a microscope to resolve fine detail. Until the last few years the practical limit of resolving power has been obtained by the use of an apochromat having a

numerical aperture of 1·40. Such a lens, properly handled, is capable of showing fine lamellar structures of the order of 140,000 lines to the inch, though it must be admitted that many published photo-micrographs of metals taken with lenses of this type exhibit nothing like this high degree of resolution. The capacity to distinguish finer detail than this might, however, throw light on structural problems still unsolved. Greater resolving power can be obtained in at least two ways—increasing the aperture of the objective or using light of shorter wave length. The first possibility was tentatively explored nearly fifty years ago when the famous Zeiss firm produced an immersion objective with a numerical aperture of 1·60, the immersion fluid being monobromonaphthalene. There were practical difficulties, however, in the use of this lens with transparent objects, as it was necessary to mount the latter on covers and slips of glass having a refractive index equal to that of the immersion fluid, if the high aperture of the lens was to be used effectively, and such slips and covers were costly and difficult to produce. These difficulties do not arise, however, with metallurgical specimens and hence it is somewhat surprising that this high-aperture objective has not been applied sooner to metallography. Its advantages for this purpose, however, were demonstrated about three years ago by Dr. F. Lucas,[1] who has since stated,[2] however, that it is not easy to use the lens, owing to trouble, some back reflections from certain lens surfaces in the objective. These reflections produced glare in the image, obscuring fine detail and reducing contrast. Meanwhile, H. Wrighton, of the Research Department, Woolwich, had been experimenting with a 2 mm. objective of the same aperture, which had been computed by R. J. Bracey, of the British Scientific Instruments Research Association and made by R. and J. Beck, Ltd. This lens is a monochromat corrected for deep blue light of approximate wave length 4395 A, and by its use lamellar structures measuring 180,000 lines to the inch have been resolved.[3] Wrighton mentions that glare has been satisfactorily reduced in this lens; his published photographs are certainly remarkably crisp and clear.

It may not perhaps be out of place to emphasise here the importance of designing a metallurgical objective so as to eliminate back reflections as much as possible. The writer of this article drew attention[4] many years ago to the advisability of giving the back lens of a metallurgical objective as short a radius of curvature as possible, so as to remove most of the troublesome glare produced by reflection from the almost plane surfaces of the back lenses of many objectives. Similar glare, though probably less marked, is also produced from reflections at air-glass surfaces within the objective. Smiles and Wrighton[5] suggest that objectives for metallurgical work should be computed with bold curvatures, so as to avoid this trouble, but they proceed: "unfortunately, very few objectives are constructed with attention to metallurgical requirements." Let us hope that the near future will see an improvement in this state of affairs.

The paper by Smiles and Wrighton deals mainly with the use of ultra-violet light as a means of increasing resolving power. These authors used a quartz monochromat of focal length 1·7 mm. and N.A. 1·25, which was corrected for light of wave-length 2750 A, and hence had an equivalent aperture 24 per cent. higher than that of the visual objective of N.A. 1·60 used with blue light. They publish a number of photo-micrographs, magnified 2500 diameters, of lamellar and sorbitic pearlite, taken with the quartz monochromat, and also with the objective of N.A. 1·60, and these show a slight but definite superiority in favour of the former lens, with which a lamellar structure of 200,000 lines per inch was resolved. The use of ultra-violet light in metallurgical microscopy is not without difficulties, but these have been very cleverly surmounted. In this respect, these authors had the benefit of help and advice from Mr. J. E. Barnard, F.R.S., whose work in all branches of photo-micrography, including the use of ultra-violet light, is well known.

As mentioned above, the photographs obtained with ultra-violet light show somewhat greater resolution than those taken with the objective of N.A. 1·60. The difference, however, is not great and if one bears in mind the advantage of the latter objective in that it can be used for visual examination, it would appear that those who wish for increased resolution of fine metal structures may find it more convenient to use this objective rather than to try and overcome the relatively immense difficulties associated with ultra-violet light.

For the great bulk of metallurgical microscopy, however, objectives of lower aperture and magnifying power supply all that is required, provided they are mounted on microscopes which allow the objectives to be used efficiently and in a convenient manner and also which permit the sample to be illuminated effectively. Thirty years ago, most available microscopes had been designed primarily for the examination, by transmitted light, of the more or less transparent objects customarily used in biology and other kindred sciences. Most of the "special metallurgical stands" then available were only slightly modified forms of this traditional model, the most important alteration being generally the provision of a focussing adjustment to the stage. For purely visual observations, it is probable that the better-designed models of this class are still at least as convenient as any other type—in the writer's opinion, more convenient than some of their modern competitors. For photo-micrography, however, these traditional models have some disadvantages, and one of the most notable developments of the last twenty years has been the gradual evolution of the modern "inverted" microscope from the original Le Chatelier model, which was designed to remove some of the inconveniences experienced with the older microscopes. This gradual evolution is well portrayed in an interesting paper by Dr. H. Freund, which appeared recently in Germany.[6] He describes and illustrates the original microscope constructed by Pellin for Le Chatelier, and those made later by such famous firms as Dujardin, Zeiss, Leitz, and Reichert. It is particularly interesting to note how the stands made by each of these firms have been modified and improved over a period of ten or twenty years.

Dr. Freund also describes some of the smaller workshop models, which combine camera, microscope, and lamp in one rigid stand, and are made by various firms. These include such well-known models as the Leitz "Panphot," the Reichert "MeF," the Busch "Metaphot," and the Zeiss "Metalliput." A somewhat similar description of such stands (and also of the larger models made by Leitz, Zeiss, and Reichert) is given by Richard Pusch in an article dealing with improvements in metal microscopy.[7] It must be admitted that these smaller models are very compact, and they appear to be well arranged. Where space is limited and total cost has to be considered very carefully, they are undoubtedly attrac-

tive to those who wish to use one microscope for both visual observations and photo-micrography, particularly under more or less standardised conditions.

Both Pusch and Freund also deal briefly with special methods of illumination, more particularly oblique or dark ground illumination and the use of polarised light. It may be recalled that a special series of objectives designed to give oblique light in all azimuths was put on the market in this country many years ago by Messrs. Chapman and Aldridge, but, so far as the writer is aware, they have not been obtainable for some time. The lenses described in the German articles are constructed in a similar manner to these earlier objectives, and doubtless perform in a similar fashion. It seems doubtful whether this type of illumination is of much use for structural investigations of metals, though it has some advantages in the examination of metal surfaces. As Professor Köster[8] points out, it may be useful in deciding whether an inclusion or a small hole is in question. The former, it may be noted, appears in its natural colour. Pusch gives one illustration (a very poor one) of the use of oblique illumination; a sample of annealed steel, containing ferrite and globular pearlite, was photographed, at a magnification of 500 diameters, in one case with vertical light and in the other case with oblique light. The detail in the latter is very poor, and would hardly encourage anyone to try this method of illumination.

The various microscopes and photo-microscopes described in these articles by Pusch and Freund are almost entirely of German and Austrian manufacture, and by comparison, one cannot help thinking that firms in these countries have shown more active efforts to meet the special needs of the metallurgists than has been the case with some of their English rivals—at least, if one may judge by the illustrations and descriptions in the current catalogues of firms in this country. There is no question, of course, about the quality of the English products—the writer has used metallurgical stands made by a well-known English firm for over thirty years, and can vouch for the high quality of their workmanship—but one could wish for a little more appreciation of the fact that a photo-microscope—microscope, camera, and illuminating system—designed primarily for use with transparent objects is not necessarily the best basis of design for a metallurgical photo-microscope, particularly if convenience and ease of working are considered.

As a contrast to these somewhat disparaging remarks, one may turn finally to a paper by H. Wrighton describing a photo-microscope,[9] constructed by R. and J. Beck, for the Research Department, Woolwich, which is intended primarily for metallurgical work, although provision is also made for dealing with transparent objects. A particular feature of this apparatus is the very complete arrangement for photography, under vertical illumination, at the very lowest powers. Comparatively large sections can be evenly illuminated and photographed. For work of this description, short focus micro-anastigmats are greatly preferable to ordinary low-power microscopic objectives; they give much flatter fields and have ample resolving power, but, if their good properties are to be adequately utilised, they require an effective system of illumination and properly designed fittings to hold them, and so often these essentials are inadequately provided!

The condenser system for high and medium powers provided in the Woolwich apparatus is also excellent, and allows "critical" illumination to be obtained with ease.

The writer has had the privilege of examining this apparatus on more than one occasion, and he can vouch for the excellence of its design and construction. He would suggest that those interested in the photo-micrography of metals at all available magnifications could study the design of this photo-microscope with advantage; they will find it equal to that of any foreign product.

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6 "Zur Geschichte des Metallmikroskops." Die Giesserei. Sept. 11th, 1936.

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Hardness of Steel at High Temperatures.

The development of nearly every instrument for static or dynamic hardness testing has been accompanied by attempts to measure hardness at high temperatures. In the first instance, attention was drawn to this property in connection with the forgeability of materials; later the discovery of high-speed tool steels which retained their hardness at raised temperatures offered a field for investigation and more recently the rapid improvement in heat-resisting steels has given rise to renewed attempts to correlate hardness with other properties at high temperatures. "Hot-hardness," with special reference to pure metals, has been discussed by J. H. Hruska.[1] He finds that for pure metals there appears to be a logarithmic relation between hardness and temperature, expressed by the following formula:—

$$\log H_2 - \log H_1 = K (t_2 - t_1),$$

and that the coefficient K is directly proportional to the absolute temperature of the melting point of the pure metal. The relationship in alloys and especially in steels is much more complex and the hardness numbers given by Hruska in the form of curves for the hot hardness of carbon steels show a far simpler relationship than could be justified by the more accurate data of F. Körber and I. B. Simonsen[2] illustrated in Fig. 1.

It has long been known that with increasing temperature the resistance to deformation of steel falls and then rises again in a certain temperature region depending on the speed of testing. A. le Chatelier[3] long ago pointed out that the influence of speed of deformation was no less important than that of temperature. The increased hardness in the "blue-heat" region has been associated by many writers with strain ageing. Non-strain ageing steels do not show this increase. It is now widely held that strain ageing

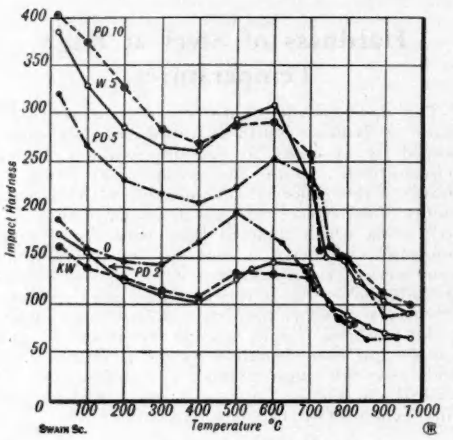
is attributable to a precipitation process, and it was to follow up this idea that R. Walzel[4] embarked on a series of tests of the "Static and Dynamic Hot Hardness of Steel."

Four steels were used having the compositions given in Table I. Static tests were made with a Widia ball 10 mm. in diameter and a load of 3000 kilos. The time to maximum load was 6 sec., and the pressure was then released. The dynamic ball test was carried out with a steel ball of 10 mm. diameter in the pendulum machine of R. Walzel[5] with a total striking energy of 349 mm-kilos. Continued heating of the test piece under test and the use of a Widia ball are unnecessary on account of the extreme rapidity of the test. The dynamic hardness number was given by the energy absorbed (*i.e.*, 349 mm-kilos. less energy of rebound) divided by the volume of the impression in cubic millimetres. The results of the tests are shown in Fig. 2. In agreement with previously published work the maximum value of the dynamic hardness lies at 500–600 deg. Cent., and that of the static hardness at 300–400 deg. Cent. But a new feature was observed in the dynamic tests, viz., the occurrence of a second maximum of the dynamic hardness at about 150–250 deg. Cent. This is not shown by the usually accepted curves, *e.g.*, those of Körber and Simonsen (Fig. 1).

That the temperature of the maximum hardness value was dependent on the duration of the test was proved by a further experiment in which the duration of the static ball test was prolonged twentyfold, the maximum load being reached by a uniform rate of loading in 120 sec., instead of 6 sec., and the pressure then released. As shown in Table II, and in accordance with expectation, at 250 deg. Cent. the prolonged static test gave the higher hardness since it offered the opportunity of a greater increase through hardening during the test, but at 500 deg. Cent. the prolonged static test gave a smaller hardness number since at this temperature softening had already set in and the long duration of the test provided additional opportunity for deformation. It is thus clear that the hardening process responsible for the first maximum in the dynamic hardness curve (at 150 deg. to 200 deg. Cent.) must be completed in a very short time, and in the static tests is masked by the slower hardening process which, in the short time

TABLE I.—*Composition and Treatment of Steels Used.*

Steel No.	C, per cent.	Si, per cent.	Mn, per cent.	P, per cent.	S, per cent.	Cr, per cent.	Ni, per cent.	Treatment.
1	0·09	—	0·43	0·011	0·031	—	—	Annealed 15 min. at 920 deg. Cent.
1A	0·08	—	0·49	0·027	0·048	—	—	,, 920 deg. Cent.
2	0·52	0·16	0·67	0·011	0·036	—	—	,, 840 deg. Cent.
3	0·83	0·19	1·00	0·047	0·050	—	—	,, 780 deg. Cent.
4	0·30	0·12	0·46	0·005	0·012	0·89	3·47	Oil hardened from 840 deg. and tempered 30 min. at 680 deg. C.



Mark ...	0	PD2	KW	PD5	PD10	W5
Carbon, per cent. ...	0·05	0·10	0·14	0·56	0·97	1·06

FIG. 1—*Dynamic Hardness of Carbon Steels at High Temperatures (Körber and Simonsen).*

TABLE II.—*Normal and Prolonged Brinell Hardness Tests (3000 kg. per 10 mm.) at 250 deg. Cent. (Steel Ball) and 500 deg. Cent. (Widia Ball).*

Composition of Steel.			Maximum load attained in			
C, per cent.	Si, per cent.	Mn, per cent.	6 sec.		120 sec.	
			250°	500°	250°	500°
0·15	—	—	130	108	140	92
0·31	0·24	0·76	141	133	154	105
0·50	0·15	0·77	185	169	208	133
0·71	0·20	0·80	232	223	245	195

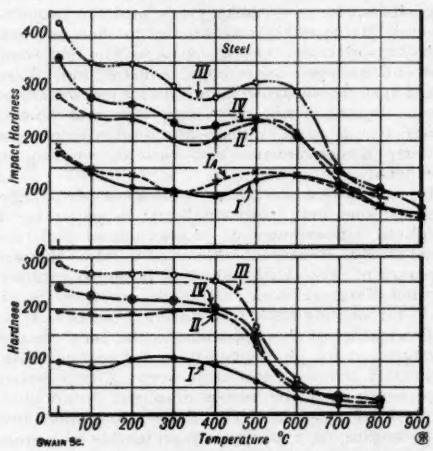


FIG. 2—*Dynamic and Static Hardness of Steels 1 to 4, Table I, at High Temperatures (Walzel).*

available in the dynamic test, requires the high temperature of 500 deg. to 600 deg. Cent. to display its maximum effect. This is illustrated by Figs. 3 and 4. A three-dimensional diagram given by the author shows that the first hardening process occurs with such great rapidity that it cannot be detected by dynamic tests at or above 300 deg. Cent. The

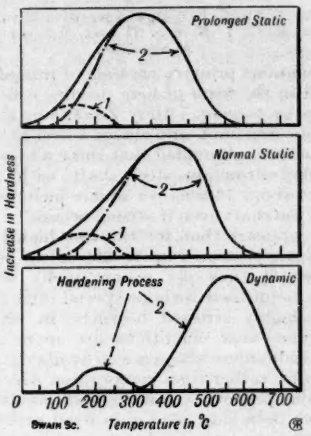


FIG. 3—*Diagram Indicating How Two Hardening Processes Would Account for the Form of the Curves in Fig. 2.*

second hardening process which causes a much greater increase in hardness is at its maximum in static tests at 350 deg. and in dynamic tests at 550 deg., while complete resoftening has occurred in the time taken by the dynamic test at 750 deg. Cent.

A study of the relation shown by steels of various compositions between hardness, temperature, and

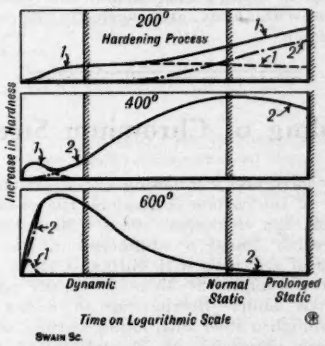


FIG. 4—*Diagram Indicating the Relative Importance of the Two Hardening Processes in Tests at Different Speeds at 200 deg., 400 deg., and 600 deg. Cent.*

duration of tests would lead to valuable practical and theoretical results. Resistance to deformation in relation to temperature and time is of importance in connection with hot working generally, and the form of the curve is also of theoretical interest. If there is an upward inflexion of the dynamic hardness curve at about 150–200 deg. Cent. as claimed by Walzel, it could no doubt be explained, as he suggests, by a hardening process distinct from that associated with strain ageing, but the nature of this process is quite unknown. Moreover, the existence of the inflexion is a new observation which will require confirmation. Notched-bar impact tests which, in a given material, usually vary inversely with the hardness do not show any similar inflexion as they rise to their maximum at 200–300 deg. Cent. The general relation between time and temperature, both for hardness and for notched-bar tests, can be explained on experimental results so far generally accepted, by the effect of time and temperature on a single strain-ageing operation, and more data will have to accumulate in the shape of the curve between room temperature and 400 deg. before it will generally be felt necessary to complicate this explanation by assuming that a second hardening process is occurring with greater rapidity at a lower temperature.

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[2] *Mitt. K.W. Inst. Eisenforschung*, 1924, 5, 21–35.

[3] *Revue de Métallurgie*, 1909, 6, 914–917.

[4] *Archiv für das Eisenhüttenwesen*, June, 1937, 1936–37, 10, 577–580.

[5] *Stahl and Eisen*, 1934, 54, 954–957.

Light Metal Bearings.

A NUMBER of recent German publications have had reference to the development of bearing metals with a magnesium or more usually an aluminium base.* Of the materials with a magnesium base, "Elektron" alloys of different types, both cast and forged, are stated to have been used recently with satisfactory results in bearings working at low pressures and at temperatures below 100 deg. Cent., efficient lubrication being essential. The aluminium-base bearing metals came into some prominence during the war, and tests were made on copper-aluminium and copper-zinc-aluminium bearings at the Royal Aircraft Establishment, the results being published in Report No. 15 of the Light Alloys Sub-Committee of the Advisory Committee for Aeronautics. The use of such alloys, however, dates back still further, as it was reported in THE ENGINEER in 1911 that bearings of the composition 92 aluminium, 8 copper, had been tested on a French railway and found satisfactory, there being no trace of wear after 50,000 miles. Experiments on the Lancashire and Yorkshire Railway showed much less favourable results, and Mr. H. N. Bassett in his recent book describes how the subsequent trials on the French railway confirmed the English experience. Mr. Bassett's book† may be consulted for detailed information on the subject of light metal bearing alloys down to a recent date.

Of the German papers which have now appeared, the most general is that of Wiechell who describes attempts made to improve the aluminium-base alloys in the light of the usual theory of the structure of bearing metals, viz., that they should contain hard

* H. Wiechall, "The Development, Construction, and Use of Light Metal Bearings," *Automobiltech. Zeit*, 1937, 40, 235; E. Vaders, "Recent Aluminium Bearing Metals," *Zeit. Metallkunde*, May, 1937, 29, 155; R. Hinzmann, "Light Metal Bearings," *Zeit. Metallkunde*, May, 1937, 29, 158; R. Hinzmann, "Light Metal Bearings," *Metallwirtschaft*, May 14th, 1937, 16, 477; M. v. Schwarz, "The Running Properties of Light Bearing Metals, with Special Reference to 'Quarzal,'" *Metallwirtschaft*, July 30th, 1937, 16, 771.

† H. N. Bassett, "Bearing Metals and Alloys," pages 367–371 (London: Ed. Arnold and Co., 1937).

particles embedded in a soft matrix. Both the size and distribution of the particles and the hardness and character of the matrix are of importance. A eutectic ground mass was found to be superior to a solid solution; therefore, additions are chosen which have a small solid solubility in aluminium, form a eutectiferous series, and, in the hyper-eutectic alloy, give primary crystals of a hard constituent, the size of which can be controlled by a "modifier," by suitable conditions of solidification or by forging of the cast alloy.

The importance of laboratory and practical full-scale tests is discussed. The laboratory tests include, in addition to the determination of the usual mechanical properties and of coefficient of expansion, wear tests and bearing tests. The wear test was made by abrasion against the steel of the shaft under different conditions of load and lubrication, wear being measured by loss of weight. The bearing test was made by rotating a bearing on a stationary steel shaft under a pressure of 200 kilos. per square centimetre, at a speed of 7 m. to 8 m. per second, and observing the rise in temperature.

These laboratory tests serve to eliminate the definitely unsuitable alloys. The practical test on full-size engine bearings further discriminates between alloys which have given satisfactory results in laboratory tests, but which may tend, for example, to heat up or seize under practical working conditions. This is mainly a lubrication question as even temporary metal-to-metal contact between shaft and bearing is liable to cause seizing, but such contact may be brought about by undue expansion of the alloy or by deformation of the support. More details as to the composition of the most recent alloys are given in the other papers mentioned. Vaders refers to the behaviour of "Alva 36," which is a lead-antimony-aluminium, with additions of copper, manganese, iron, &c., the composition of which was not given. Hinzmann mentions a variety of alloys, *e.g.*,

(*a*) "Borotal Z 7" Cu 3 to 4, Fe up to 2, Pb and Zn up to 3 per cent.; Al remainder
"Borotal D 4 and D 5" The same as Z 7 without zinc
"Borotal FZ 17*a*" ... The same as Z 7 without copper

These all contain graphite to the extent of about 0·1 per cent. of carbon. They vary in hardness and in suitability for different conditions of use.

(*b*) "Alugir" ... Cu 3, Zn 0·8, Ni 1 to 1·5 per cent.; Al remainder
"Chromet" Si 10 per cent.; Al remainder
"RR 56" ... Cu 2, Si 0·6, Mg 0·8, Fe 1·4, Ni 1·2, Ti 0·1 per cent.; Al remainder Duralumin and Silumin.

These are mentioned as giving good results as bearings running on hardened shafts.

(*c*) "Quarzal 5" Cu 5, metals of the iron group 1, refining additions 0·5 per cent.; Al remainder

Provided that this is diamond-machined, and allowance is made for its high coefficient of expansion, it is stated to present no difficulties.

(*d*) "KS 280" ... Si 21 to 22, Co 1·2, Cu 1·5, Mn 0·6, Ni 1·5, Mg 0·5 per cent.; Al remainder

This is a comparatively hard alloy (Brinell hardness 120) of low coefficient of expansion, which requires accurate machining, a hardened polished shaft, and very clean oil. In tests under satisfactory conditions the wear is given as only about one-seventh of that of a lead and tin-bronze. A special process is used to ensure a fine and uniform distribution of the free silicon particles conferring on the alloy a structure like that of the white metal bearings.

(*e*) "KS 13" ... Sb 6 to 8 per cent.; Al remainder

The Brinell hardness of this alloy is about 40, and it must be used in layers about 1 mm. thick in a supporting shell which may be made of KS 280 or duralumin.

(*f*) Alloy X ... Fe 6, Mg 0·5 per cent.; Al remainder
Alloy Z ... Ni 6·5, Ti max. 0·5 per cent.; Al remainder

Alloy X contains primary crystals of iron-aluminium compound in the form of long needles which can be broken up by forging. Alloy Z contains the nickel-aluminium compound, and shows a desirable bearing metal structure. Provided that these alloys are used on a nickel-chromium steel shaft with a tensile strength of about 75 tons per square inch, no damage is done to the shaft even if seizing occurs.

It thus appears that for the less highly stressed bearings with pressures up to 200 kilos. per square centimetre (1¼ tons per square inch), aluminium alloys can be utilised without special difficulty. For the more highly stressed bearings, in which temperature and wear conditions are more severe, a variety of aluminium alloys is also available. In these it is structure rather than composition which confers good running properties, but the multiplicity of types which have been suggested is in itself proof that no alloy of outstanding merit has been discovered.

The recent interest in these alloys in Germany is probably mainly due to the desire to utilise all metals in the most economic and effective way, coupled with a shortage of tin. The development of a light metal bearing, in place of lead-bronze, for aircraft engines is also an attractive proposition. A fair measure of success has attended the development of aluminium-base bearing metals, but in practice this can only be maintained by the exercise of extreme care in the design of the bearing, the nature of the working surfaces, the play allowed between journal and bearing in relation to heating effects, and the coefficient of expansion of the alloy, and especially the conditions of lubrication.

Scaling of Chromium Steels.

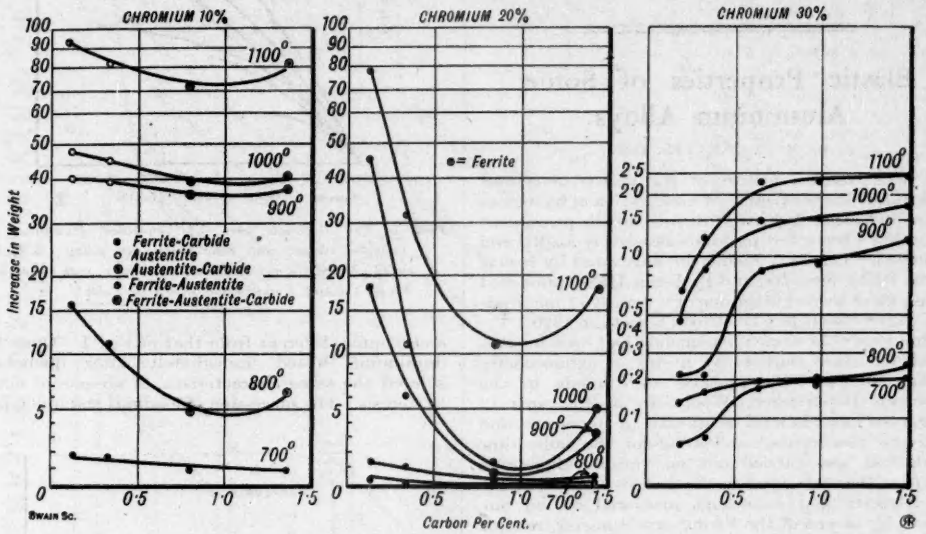
THERE have been conflicting statements about the influence of the carbon content on the resistance to scaling of high chromium steels. M. Schmidt and O. Jungwirth[1] found no difference in the amount of scaling of six steels with carbon 0·12 to 1·03 per cent. and chromium 21·45 to 31·88 per cent. after twenty-four hours' heating in air in an electric muffle, though a steel with higher carbon and somewhat lower chromium (1·72 and 20·30 per cent. respectively) showed about 50 per cent. more scaling than the others at 1000 deg. Cent. and above. The method employed, viz., to measure the loss in weight of the specimen after the scale had been detached by blows of a mallet, was, however, one which might lead to false conclusions on account of the variable degree of adhesion of the scale. In steels with 12·5 per cent. of chromium and 1·38 to 2·38 per cent. of carbon, J. P. Gill[2] found the best scaling resistance in the steel with the lowest carbon content of the series. According to K. Roesch[3] the resistance to scaling of a series of high chromium cast alloys

was more or less independent of the carbon content, but increased approximately in proportion to the chromium content and reached a maximum with 32 per cent. of chromium when the material was still scaleproof at 1200 deg. Cent. E. Houdremont [4] found in steels with 10 to 18 per cent. of chromium and 0·4 to 2·0 per cent. of carbon a maximum resistance to scaling at 1 per cent. of carbon.

In order to investigate the question further and clear up some discrepancies, scaling tests on steels of different chromium and carbon contents have been carried out at Aachen by W. Oertel and W. Landt.[5] The steels were made in a 2·5 kilos. acid high-frequency furnace from Armco iron, a steel containing carbon 0·11, silicon 0·3 and manganese 0·5 per cent., metallic chromium, and Swedish white iron. They contained 0·10 to 0·27 per cent. silicon and 0·12 to 0·25 per cent. manganese. The chill-cast ingots were forged into bars 18 mm. by 15 mm. in section. From these specimens, 20 mm. by 15 mm. by 12 mm., were taken for scaling tests, and ground smooth on all sides. The tests took place in air or in flue gas obtained by burning coal gas, and containing 0·5 to 0·7 per cent. of oxygen and 7 to 8 per cent. of carbon dioxide. In the experiments in air the specimens were heated in a tube furnace plugged at the end with a wad of asbestos to prevent air currents. In the other series the flue gas was led into the testing furnace, but the specimens were in no case exposed to any flame. All the specimens of the same chromium but variable carbon content were tested at the same time, and the degree of scaling was determined by measurements of the increase in weight. In Figs. 1 to 3 these weight increments are plotted on a logarithmic scale and the structural characteristics, deduced from the work of W. Tofaute, C. Küttner, and A. Büttinghaus[6] or of V. N. Krivobock and M. A. Grossmann,[7] are also indicated.

In the 10 per cent. chromium series, Fig. 1, there is no marked effect of carbon content at 700 deg. Cent., at which temperature all four steels consist of ferrite and carbide. At 800 deg. Cent. there is a maximum resistance to scaling at about 1 per cent. of carbon. Steels 3 and 4 contain austenite as well as ferrite and carbide. At 900 deg. and 1000 deg. Cent., when steels 1 and 2 consist of austenite, and 3 and 4 of austenite and carbide, the differences are less, but they reappear at 1100 deg. Cent., when steel 1 again contains ferrite. In flue gas the scaling of these steels was about three times as great as in air, but it was independent of the carbon content, all four steels giving the same results.

In the 20 per cent. chromium series, Fig. 2, scaling was about the same for all four steels (Nos. 5 to 8) at 700 deg. and at 800 deg. Cent., the structure of each being ferrite and carbide. At 900 deg. Cent. austenite appears in steels 7 and 8, which then show very little scaling in comparison with the low carbon steels of the same series. This effect was intensified



FIGS. 1 TO 3—*Influence of Carbon Content on the Resistance to Scaling of Chromium Steels at High Temperatures in Air.*

Fig. 3. The structure of this series consists of ferrite and carbide at all temperatures and carbon contents. Accordingly, these steels show quite a different form of curve, in which resistance to scaling decreases with increasing carbon content, *i.e.*, with increasing proportion of carbide, whereby the chromium content of the ferrite is correspondingly reduced. The predominant effect of increase in chromium content may be noted by comparing the behaviour of the ferrite-carbide steels in each series. The weight increments

at 1000 deg. and 1100 deg. Cent. Curves for scaling in flue gas also showed a minimum amount of scaling at about 1 per cent. of carbon, but a much smaller relative difference between different steels and a much greater amount of scaling than in air, *e.g.*, steels 5 to 8 at 1100 deg. Cent. gave weight increments of about 160, 150, 120, and 160 g/hm² respectively. These steels were, however, considerably more resistant to flue gas than the 10 per cent. chromium series, the corresponding figure for which at 1100 deg. Cent. was 250 g/hm².

The experimental data obtained with the steels containing 30 per cent. of chromium are shown in

of all the 30 per cent. chromium steels were very low.

The work of Oertel and Landt serves to direct attention to the fact that austenite of a given composition is intrinsically more resistant to scaling than ferrite and carbide of the same composition, and that in comparing the results of scaling tests on chromium steels with different carbon contents, due allowance must be made for differences in constitution which accompany differences in temperature and in composition.

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Elastic Properties of Some Aluminium Alloys.

By R. H. GREAVES, D.Sc.

A WELL-KNOWN feature of the plastic behaviour of metals is the occurrence of a succession of hysteresis loops increasing in area with increase in permanent extension when a test piece is successively loaded and unloaded. This may readily be illustrated by means of the Dalby recorder, and Professor Dalby published examples of looped diagrams of a variety of materials in a paper read before the Royal Society in 1920.[1]

Previously the same phenomenon had been investigated by other workers by means of extensometer measurements.[2] Observations were made in the Research Department, Woolwich, of the area of hysteresis loops in steel as affected by overstrain and recovery treatments,[3] and at about the same time work was also carried out on various aluminium alloys. The tests were made on test pieces 0·792in. in diameter and extensions measured on an 8in. length by means of the Ewing extensometer, reading to 2×10^{-5} inch. The usual behaviour of aluminium or of an aluminium alloy is as shown in Fig. 1. After the yield point had been exceeded the area of the loop produced by unloading and reloading increased with the amount of permanent set, but at a diminishing rate. The diagram, Fig. 2, shows the relative areas of the loops plotted against the permanent set for a variety of aluminium alloys in the annealed or hot-rolled condition. For a given width of loop the area will, of course, increase with yield point, *i.e.*, with the range of stress over which the loop is taken. The unit of the ordinates expressing relative area of loop corresponds to a hysteresis loss of one two-thousandth of an inch-ton per cubic inch.

In the paper by Professor Dalby referred to above there was one exception to the typical looped diagram. This was shown by a material, described as an aluminium alloy, in which removal and reapplication of the load did not produce a looped area, and was thus apparently effected without incurring any elastic hysteresis losses. This behaviour was known not to be typical of all aluminium alloys, and it was ascertained that the alloy in question was duralumin rod of tensile strength 23·7 tons per square inch. At a much later date W. Schmidt[4] published diagrams for elektron and duralumin. Elektron gave considerable and increasing hysteresis as the permanent set increased; duralumin showed no hysteresis loop at 0·2 per cent. permanent set and only a very small loop at 0·4 per cent.

The change in loop area with increase in extension of quenched and aged duralumin is shown in Fig. 3,

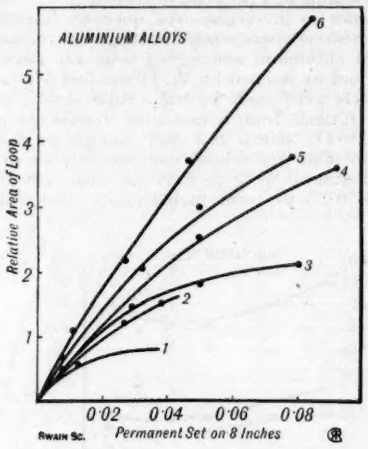

FIG. 2—*The Materials were*:—1, *Annealed Aluminium*; 2, *Annealed 15 per cent. Zinc-Aluminium Alloy*; 3–6, *Hot-rolled Nickel-Copper-Aluminium Alloys with Increasing Nickel Content, Composition shown in Table I.*

and is quite different from that of Fig. 1. Quenched duralumin, tested immediately after quenching, showed the same characteristic of absence of elastic hysteresis. The succession of loadings and unloadings

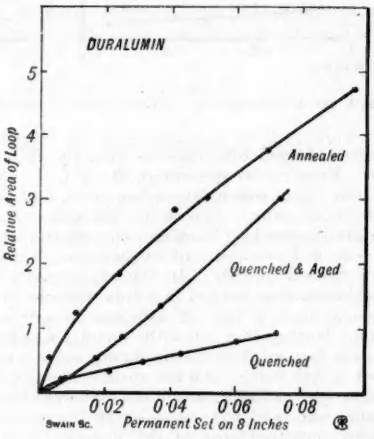

FIG. 6

shown in Fig. 4 occupied about one hour from the time of quenching. On the other hand, annealed duralumin (Fig. 5) behaved exactly like the aluminium alloys referred to in Fig. 2. The increase in loop area for duralumin in different conditions is shown in Fig. 6. The great difference between the "quenched

ELASTIC PROPERTIES OF ALUMINIUM ALLOYS

COPPER-NICKEL-ALUMINIUM ALLOY Cu 2 1% Ni 5 3%
Annealed
DURALUMIN Quenched & Aged
DURALUMIN Annealed
"Y" ALLOY Annealed
DURALUMIN Quenched & Tested Immediately
"Y" ALLOY Quenched & Aged
Tons Per Sq. Inch
Inches Extension on 8 Inches

FIG. 1 FIG. 3
FIG. 4 FIG. 5
FIG. 7 FIG. 8

and aged" and the "annealed" conditions is shown when the permanent set is still low, *e.g.* before it reaches 0·02 in. on 8 in. When the permanent set reaches about 1 per cent. (0·08 in. on 8in.), the loop area of the quenched and aged specimen is approaching that of the annealed material, although the range of loading and unloading is about twice as great, so that the width of loop, or the hysteresis per unit increment of load, is still very much less in the quenched and aged than in the annealed duralumin.

That the behaviour was related in some way to the structure developed by quenching and ageing was shown by the fact that "Y" alloy behaved similarly (Figs 7 and 8). In this case a small but definite loop was observed after the test piece had received a very small amount of permanent set, but the loop area did not increase appreciably in the quenched and aged specimen as it did in the annealed material. Complete elimination of the hysteresis loop was not obtained in any copper-rich alloys susceptible to precipitation hardening though the results indicated that the increase in loop area was much slower for alloys in the precipitation hardened condition than for the same material annealed.

TABLE I.—*Young's Modulus of Aluminium Alloys.*

Condition of rolled rod as tested.	Young's modulus.		
	Tons per sq. inch.	Lb. per sq. inch.	Kg. per sq. mm.
Commercial aluminium (99·5 per cent.)—			
Annealed, overstrained 0·2 per cent.	4320	$9·7 \times 10^6$	$6·8 \times 10^3$
Annealed, overstrained 3·5 per cent. and aged	4450	10·0	7·0
Cold drawn	4460	10·0	7·0
Zinc-aluminium alloy (Zn 15 per cent.)—			
Annealed, 350 deg. Cent.	4660	10·4	7·3
Cold drawn	4400	9·85	6·9
Nickel-copper-aluminium alloys, hot rolled—			
Cu 2·0, Ni 1·1 per cent	4570	10·2	7·2
2·0 2·2 ,,	4640	10·4	7·3
2·1 3·7 ,,	4780	10·7	7·5
2·1 5·3 ,,	5000	11·2	7·9
Duralumin (Cu 4·3, Mg 0·7, Mn 0·4, Fe 0·5, Si 0·4)—			
Annealed	4700	10·5	7·4
Quenched and aged	4660	10·4	7·35
Then overstrained 1 per cent. and aged	4620	10·3	7·3
"Y" alloy (Cu 4·0, Mg 1·5, Fe 0·6, Si 0·13, Ni 1·9)—			
Annealed	4900	11·0	7·7
Quenched and aged	4800	10·75	7·55
Then overstrained 1 per cent. and aged	4680	10·5	7·4

These results were not published at the time they were obtained, as it was hoped to clear up the reason for the apparent absence of elastic hysteresis in some of the age-hardened alloys. As, however, this has not been done and the question of damping is now receiving much attention, they are now put forward in the hope that they may assist in drawing attention to certain data which require consideration in any theory of damping of aluminium alloys which may be suggested.

YOUNG'S MODULUS OF ALUMINIUM ALLOYS.

In the course of this work some fairly reliable figures for Young's modulus of aluminium alloys were obtained with the Ewing extensometer on 8in. gauge lengths. These are given in Table I. The addition of alloy elements to aluminium increases the modulus, copper and nickel being much more effective in this respect than an equal percentage of zinc. The modulus of hard-drawn duralumin wire (18 S.W.G.) was found by Searle's method to be $10·0 \times 10^6$ lb. per square inch, and that of the same wire quenched, overstrained 1 per cent. and aged, $10·2 \times 10^6$ lb. per square inch. It seems that in general cold working or overstraining followed by ageing slightly reduces the modulus. Commercial aluminium appeared to be an exception, but in this case the limit of proportionality of the annealed metal was so low that it was difficult to obtain an accurate determination of the modulus.

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Macrostructure and Mechanical Properties.

A SHORT paper by K. Kornfeld* on "The Influence of Casting Structure on the Mechanical Properties of Forged Steel" contains interesting data on the relation between the primary structure of the ingot persisting after hot working, and the properties of the finished product. The steel used was a nickel-chromium-molybdenum steel of the composition shown in Table I. An ingot 240 mm. square was forged down to 80 mm. square, and the billet so obtained was air hardened from 850 deg. Cent., tempered for three hours at 650 deg. Cent., cooled to 300 deg. Cent. over a period of eighteen hours, and then cooled in air. The billet was cut into twenty-five longitudinal sections, five along each side, and these were again tempered at 650 deg. Cent.

In carrying out work such as this, it is, of course, necessary to distinguish clearly between lack of uniformity due to mass effect and variations due solely to differences in structure. Hence, the final heat treatment must always be applied to specimens of the same size after they have been cut from the larger billets. Accordingly, specimens obtained as described above were held for forty-five minutes at temperatures varying by 10 deg. intervals from 795 deg. to 845 deg. Cent., half were cooled in air and the other half cooled to about 600 deg. in air and then quenched in oil. All specimens subsequently received a low-temperature treatment at 180 deg. Cent. for two hours in an oil bath for the relief of internal stress. Many numerical results are given in the paper, but those abstracted in Table I are typical and illustrate the difference in properties of material situated at different positions in the cross section of the billet. The macrostructure of the outer zone (positions A, B, C) was dendritic, that of the inner region (positions D, E) equiaxial. There was little difference in the hardness, yield point, and tensile strength at different positions, though the specimens from the outer zone were generally slightly harder; but the test pieces from the inner region almost invariably showed some-

* *Stahl und Eisen* August 5th, 1937, 57, 870.

what higher elongation and reduction of area, together with an impact figure 20 or 25 per cent. higher than that of the outer material. Similar indications of the influence of position were obtained when other low-temperature treatments were applied.

In a further series of experiments, the billets were again forged to 46 mm. square at three different forging temperatures, viz., 1100 deg. to 880 deg., 1050 deg. to 830 deg., and 1000 deg. to 780 deg. Cent. Each bar was cut into three, one being cooled in air, another in dry sand, and the third in hot sand, cooling at 25 deg. per hour from 650 deg. Cent. The bars were cut into nine longitudinal sections, three along each side. These specimens were air hardened from 825 deg. Cent. and tempered at 180 deg. Cent. for two hours. The results shown in Table II are typical of those given in greater detail in the original. The forging temperature and rate of cooling after forging exert only a small influence on the tensile properties of the finally treated steel, but the influence of position in relation to macrostructure is again evident. The yield point and tensile strength were lower and the ductility greater in the middle, equiaxial region. Owing to the additional work done on the steel the tensile strength tended to be greater and the elongation lower than shown in Table I.

In translating his results into practice the author observes that if a nickel-chromium steel shows a low impact figure without the occurrence of temper-brittleness it is generally on account of the presence of a well-preserved dendritic structure. An example is given of variation in dendritic structure in a nickel-chromium-molybdenum steel, which is held to be the sole cause of a difference of impact figure of 4·7 and 10 mkg./cm.² with a tensile strength of about 57 tons per square inch in both cases.

The paper does not discuss the relative properties of the steel along and across the dendrites. All the tests in the outer zone were more or less across the dendrites, but it is reasonable to suppose that the orientation of the dendritic crystals with respect to the axis of the test piece will have a considerable effect on the properties and so also will the nature of the interdendritic segregate. This may account for some irregularities shown by the detailed results in the relative value of tensile strength of specimens taken from the corner and from the middle of the edge. The conclusion, however, appears to be well established that for a given value of the tensile strength the equiaxial macrostructure is characterised by the highest elongation, reduction of area, and notched bar impact value, and should therefore be the condition to be sought in hard materials required to show the greatest possible degree of toughness.

TABLE I.—*Mechanical Properties of Nickel-Chromium-Molybdenum Steel in relation to the Position of the Specimen in the Cross Section.*

C 0·30, Mn 0·60, Si 0·28, Ni 4·60, Cr 1·10, Mo 0·18 per cent.

Treatment	Air hardened from 825 deg. Cent., tempered 2 hours at 180 deg. Cent.			Cooled 825 deg. to 600 deg. Cent. in air, then in oil, tempered 2 hours at 180 deg. Cent.		
Position	Corner A.	Edge B* or C.	Inside D or E.*	Corner A.	Edge B* or C.	Inside D or E.*
Brinell hardness number	477	477	461	477	477	477
Yield point, tons per square inch	94·1	93·6	92·2	96·5	95·5	92·4
Tensile strength, tons per square inch	110·7	110·6	107·4	110·0	110·0	105·7
Elongation, per cent. ($l=4\sqrt{d}$)	15·3	15·0	15·0	14·6	15·0	15·6
Reduction of area, per cent.	49·1	49·2	50·9	48·5	50·7	50·8
Impact figure, mkg./cm.²	3·2	3·3	3·7	3·5	3·7	4·1

* Impact test only. Round Izod test pieces.

TABLE II.—*Mechanical Properties of Nickel-Chromium-Molybdenum Steel in relation to the Position of the Specimen in the Cross Section. Steel of composition shown in Table* I, *Forged at* 1100–880 *deg. Cent., air hardened from* 825 *deg. Cent. and tempered for two hours at* 180 *deg. Cent.*

Cooled after forging	In air.			In dry sand.			Slowly in hot sand.		
Position	Corner.	Edge.	Inside.	Corner.	Edge.	Inside.	Corner.	Edge.	Inside.
Yield point, tons per square inch	101·4	96·0	93·6	96·0	103·7	93·1	91·7	96·1	88·0
Tensile strength, tons per square inch	117·4	117·4	108·2	114·0	117·8	102·9	117·4	114·2	107·3
Elongation per cent. ($l=4\sqrt{d}$)	12·8	13·6	14·0	13·6	13·6	14·8	13·0	13·8	14·0
Reduction of area, per cent.	50·8	50·6	50·8	52·0	50·3	52·2	50·5	53·0	53·0
Impact figure, mkg./cm.²	2·1	3·1	3·7	2·8	3·3	3·6	3·5	2·8	4·0

The Diffusion of Hydrogen through Steel.

THE mechanism of the diffusion, through steel, of hydrogen produced by chemical or electro-chemical means in the process of pickling has given rise to much discussion. There are points of similarity and of difference between the hydrogen absorption which occurs on pickling at temperatures only slightly above atmospheric and that which may occur during the annealing of steel in a reducing atmosphere at high temperatures. In both cases the hydrogen absorbed is in the atomic condition, produced as such on the surface of the sheet during pickling, or formed by dissociation or decomposition of hydrogen compounds on the surface of the metal

in high temperature operations. A question which has given rise to many different expressions of opinion is that of the path by which diffusion takes place, whether penetration is along the grain boundaries or through the crystal lattice of the grains themselves. The answer has frequently been supplied on the evidence of microstructure, based on the examination of materials which have absorbed hydrogen at high temperatures. The examination of such material often suggests that the path of the gas has been along the grain boundaries or the cleavage planes, but deductions arrived at by this method are frequently misleading. The observed changes in structure are ultimately due to reactions with absorbed hydrogen, but do not necessarily imply that the presence of hydrogen was limited to those regions in which the changes are evident.

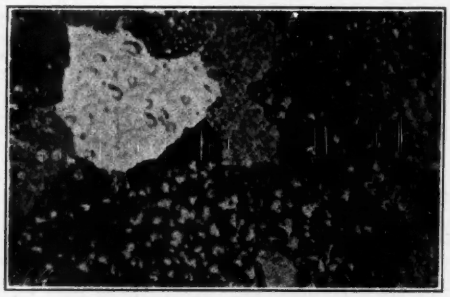

FIG. 1—*Appearance of Gas Bubbles on the Surface Etched and Coated with Canada Balsam After the Opposite Face had been Pickled for 90 Minutes.* ×75.

Thus intercrystalline failure or reduction of cohesion on cleavage planes is not necessarily a proof that hydrogen travels preferentially by these paths. The existence of grain boundary constituents, which will react with hydrogen, may intensify the mechanical effects of absorbed hydrogen in the grain boundary region though the general distribution of the gas may be uniform. Intercrystalline fissures formed by the prolonged heating of mild steel, exposed to the action of hydrogen under pressure, or to ammonia gas, are probably due to reaction of the hydrogen with carbides situated in the grain boundaries.

In spite of the attractive idea that the grain boundary would, because its atomic spacing is greater than that of the crystal lattice, offer an easier path to the diffusing gases, there is no evidence whatever to support the idea that the diffusion of hydrogen in steel occurs preferentially along the grain boundaries. Many years ago C. A. Edwards[1] found that hydrogen passed quite as readily through a specimen of iron consisting of only one crystal in the area acted upon by the pickling acid as through the same specimen recrystallised by reheating to above Ac_3, and then containing the usual structure of small crystals. From accurate experimental measurements made with hydrogen on a single-crystal tube of iron in which the measured rate of diffusion over a range of temperature up to nearly 800 deg. Cent. was found to be the same as with a similar tube having 100 grains to the square millimetre, C. J. Smithells and C. E. Ransley[2] concluded that the presence or absence of grain boundaries had no effect on the process of diffusion. In experiments of a lesser degree of quantitative accuracy, W. Baukloh and H. Guthmann[3] found that the amount of hydrogen diffusing through steel tubes increased with increase of grain size suggesting that the grain boundaries were an obstacle rather than a help to the diffusion of hydrogen. These results were, however, complicated by the possible effect of the distribution of the carbide constituent, increasing amounts of which were found to decrease the rate of diffusion of hydrogen through the steel.

The question of the influence of grain size on the rate of diffusion of hydrogen through steel has recently been examined afresh by W. Baukloh and W. Retzlaff[4] in a paper on "Hydrogen Penetration of Steel in Electrolytic Pickling." The experiments were carried out with the same apparatus as was used by W. Baukloh and G. Zimmermann.[5] The rate of hydrogen diffusion in c.c. per hour was practically identical in specimens of sheet steel, 0·2 mm. thick, having grain size in the ratio of about 1 to 20,000. This fully confirms the conclusion that hydrogen does not travel preferentially along the grain boundaries. Baukloh and Retzlaff, however, devised a more direct demonstration that the hydrogen was passing through the crystal itself. They took several sheets containing, in part,

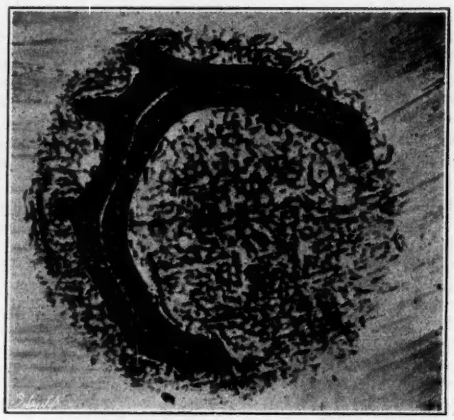

FIG. 2—*Appearance of the Surface Coated with Canada Balsam Opposite a Spot Etched with Nitric Acid on the Pickled Surface. The appearance of gas bubbles was quickly succeeded by the formation of a large blister which has broken away before the surrounding region is appreciably affected.*

grains which were many times larger than the thickness of the sheet and having etched one side and coated it with Canada balsam, they subjected the other side to electrolytic pickling for 1½–2 hours. The arrival of the hydrogen through the sheet was marked by the appearance of small bubbles in the Canada balsam, and from the arrangement of the bubbles it was clearly evident that the hydrogen can diffuse directly through the crystal grain and does not migrate preferentially along the grain boundary. The same observation was confirmed when the sheet was pickled by an ordinary acid bath without electrolytic attack (Fig. 1).

The same method was used with interesting results to demonstrate the effect of surface condition on hydrogen absorption in pickling. A small part of the surface of the sheet to be pickled was first etched for some seconds with dilute nitric acid. The sheet was then subjected to electrolytic pickling on one side (the other being coated as before with Canada balsam), whereupon a large blister quickly appeared

in the Canada balsam at a position corresponding with the etched spot (Fig. 2). The speed with which the blister was formed was an indication of the greatly increased rate of absorption and diffusion at the etched surface. It may be recalled that Smithells and Ransley[2] found that this etching treatment increased the rate of diffusion ten times.

In this case, also, ordinary pickling with sulphuric and hydrochloric acids gave the same results as electrolytic pickling. In the case of a mechanically deformed surface this was not so. When the sheet was scratched with a hard file the appearance of the Canada balsam on the "exit" side was different according to whether the scratched side was being subjected to ordinary or electrolytic pickling. The region opposite the file marks was nearly free from bubbles when electrolytic pickling was employed, but crowded with bubbles when the sheet was undergoing ordinary acid pickling. This, the authors point out, shows that some kind of mechanical roughening of the surface hinders the absorption and diffusion of hydrogen in electrolytic pickling while it accelerates it in ordinary acid pickling. The checking of the passage of hydrogen in electrolytic pickling is only temporary. After a time it begins, accompanied by the appearance of bubbles in the Canada balsam at positions corresponding to the scratches on the other side.

The method of Baukloh and Retzlaff constitutes a useful way of indicating the position of regions of preferential diffusion in a non-uniform sheet, and affords a striking demonstration in support of the conclusion that diffusion of hydrogen through steel sheet in the pickling process does not take place along the grain boundaries, but through the actual crystal grains of the steel.

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[5] *Archiv für das Eisenhüttenwesen*, 1935–36, 9, 459. See also THE METALLURGIST (August, 1936), 1935–36, 10, 157.

Iron-Manganese-Silicon and Iron-Manganese-Sulphur.

THE study by Vogel and Döring of the effect of carbon on the constitution of iron-manganese alloys[1] has been followed by further similar work by R. Vogel and H. Bedarff[2] on "The Iron-manganese-silicon System," and by R. Vogel and W. Hotop[3] on "The Iron-Iron Sulphide-Manganese Sulphide-Manganese Equilibrium Diagram." In both cases the ternary systems show a considerable degree of complication, and the diagrams now published must be regarded as, to some extent, schematic, and as preliminary guides to a more intensive investigation of regions of composition which on practical or theoretical grounds are shown to be important. The constitution of the iron-manganese-silicon alloys has been deduced in general outline from that of the corresponding binary and pseudo-binary systems, and has been checked at numerous points by the help of thermal and microscopical studies of alloys of selected composition. Sections across the ternary diagram at 15 and at 30 per cent. of manganese are shown in Figs. 1 and 2, in which the use of the following abbreviations may be noted :—D = Mn_5Si_3, E = MnSi, F = FeSi, G = Fe_3Si_2, H = 4 Fe_3Si_2, Mn_5Si_3 or $Fe_{12}Si_{11}Mn_5$; E F and D H represent solid solutions of the respective compounds, while M C refers to a ternary solid

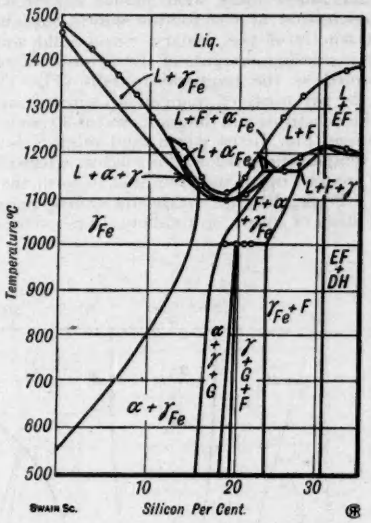


FIG. 1—*Section Across the Iron-Manganese-Silicon Diagram at 15 per cent. of Manganese.*

solution of Mn_5Si_3. The constitution of the alloys at 400 deg. Cent. is shown in Fig. 3, which was constructed by the aid of observations made on alloys which had been cooled between 1500 deg. and

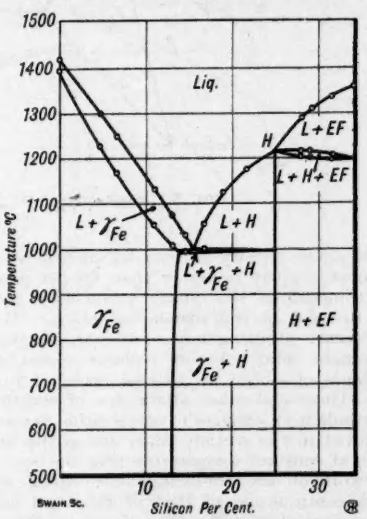


FIG. 2—*Section Across the Iron-Manganese-Silicon Diagram at 30 per cent. of Manganese.*

1000 deg. Cent. (where the crystallisation and transformation processes mainly occur) at a rate of about 1 deg. to 1·5 deg. Cent. per minute.

In the binary iron-silicon alloys the composition at

which the γ phase disappears is about 1¼ per cent. of silicon. As the manganese content increases this limit is extended. With 8 per cent. of manganese some γ-iron persists at the solidus temperature when 14 per cent. of silicon is present; and in the 15 per cent. manganese alloy with similar silicon content the constitution at the solidus shows the alloy to consist wholly of the ternary γ-iron solid solution, though increasing amounts of the α-iron solid solution are formed as the temperature falls (Fig. 1). At about 28 per cent. of manganese α-iron disappears from the constitutional diagram, and at 30 per cent. of manganese (Fig. 2) the γ-iron solid solution is stable at all temperatures below the solidus, whatever the silicon content up to that required to form the compound 4 Fe_3Si_2, Mn_3Si_2. In the manganese-rich alloys the addition of silicon up to about 10 per cent. tends to preserve the ternary γ-iron solid solution when the manganese content is higher than 65 per cent., at which composition the binary γ-iron solid solution ceases to exist in iron-manganese alloys. Beyond 75 per cent. of manganese, however, the ternary α-manganese solid solution replaces γ-iron at all percentages of silicon up to that required to form Mn_3Si. These and other alterations of constitution which result from changes of composition can readily be followed in the section taken across the ternary diagram at constant temperature (Fig. 3).

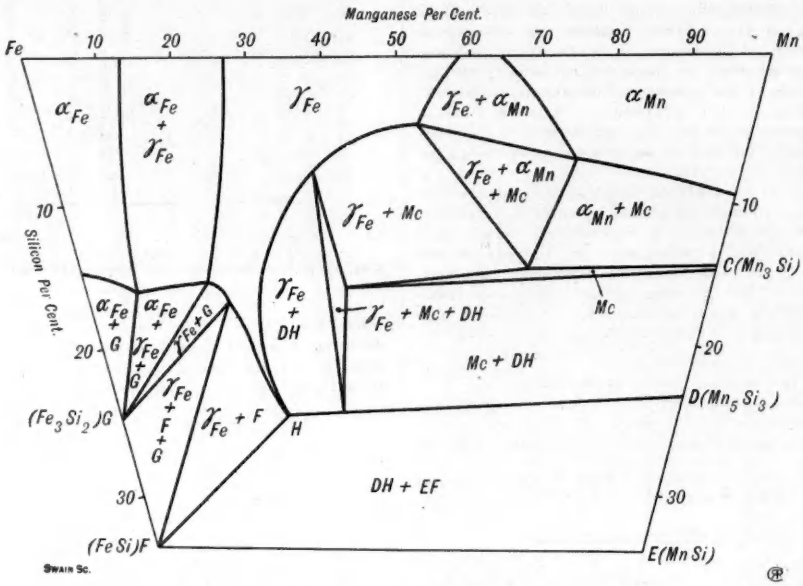


FIG. 3—*Constitution of the Iron-Manganese-Silicon Alloys at 400 deg. Cent.*

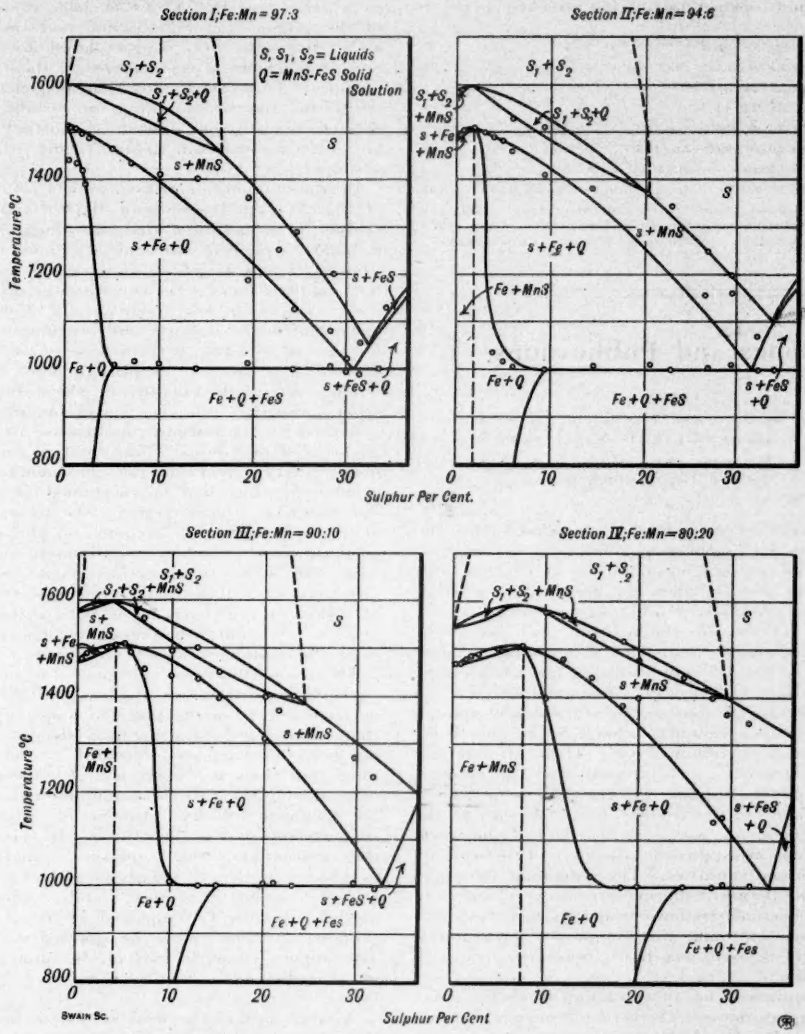
The work on the iron-manganese-sulphur system forms a continuation of that of R. Vogel and H. Baur,[4] published in 1933. Since that time the extent of the miscibility gap in the liquid condition has been re-investigated by F. Körber and W. Oelsen,[5] O. Meyer and F. Schulte,[6] and by H. Wentrup,[7] whose work is reviewed by the authors with the object of selecting the most trustworthy data. Vogel and Hotop also accept the conclusions of previous investigators of the related binary systems,[8] with the modification that for the melting point of FeS they use 1191 deg. Cent. instead of 1173 deg., and the consequent correction of the FeS-MnS eutectic temperature (which is known to be 11 deg. lower than the melting point of pure FeS) to 1180 deg. Cent.

The experimental investigation of the diagram made by Vogel and Hotop consisted of thermal and microscopical examination of four series of alloys, made from Krupp's "W.W." iron, "Thermit" manganese and pure sulphur, and forming four sections across the diagram with constant iron to manganese ratio and increasing sulphur content. These were (Figs. 4 to 7), section 1, iron to manganese = 97 to 3, fourteen alloys containing 0·5 to 35 per cent. of sulphur; section 2, iron to manganese = 94 to 6, fifteen alloys containing 1 to 35 per cent. of sulphur; section 3, iron to manganese = 90 to 10, fifteen alloys containing 1 to 33 per cent. of sulphur; section 4, iron to manganese = 80 to 20, seventeen alloys containing 1 to 35 per cent. of sulphur.

The iron-manganese-sulphur system has in the past proved to be a particularly difficult one to investigate, mainly on account of the action of the molten materials on crucibles, involving the introduction of impurities into the alloys, and on pyrometer sheaths leading to damage of thermo-couples and consequently to erroneous results in thermal analysis. These sources of error were guarded against by Vogel and Hotop, who used melting appliances and protective tubes of sintered corundum in place of the Pythagoras tubes previously employed. The "Degussa" crucibles and tubes are specially recommended. The specimens were melted in argon instead of in nitrogen. A series of chemical analyses was carried out in order to establish the equilibrium proportions in the liquid state, and the melting experiments were frequently repeated to

ensure that the data obtained were entirely trustworthy.

Solidification of these alloys occurred under the conditions imposed by the existence of two maximum lines of three-phase equilibrium, both of which in projection practically coincided with the diagonal drawn across the system from Fe to MnS. The upper one at 1600 deg. Cent. corresponds to a maximum temperature in the equilibrium of an iron-mangangese solid solution, a sulphur-rich liquid and an almost sulphur-free (metallic) liquid; the lower one (at 1510 deg. Cent.) to a maximum temperature in the equilibrium of an iron-manganese solid solution, an iron sulphide-manganese sulphide solid solution, and a sulphur-rich liquid of eutectic composition. The composition of the alloy in relation to the position of these maximum lines, *i.e.*, its composition in relation to the maxima on the curves in the series examined, determined its constitution and structure. The only sulphide present when the sulphur content was below that at which the maximum at 1510 deg. Cent. occurred was MnS. With more sulphur FeS was found in solid solution with MnS in the regions indicated by Q in the diagrams (Figs. 4–7). At 1000 deg. Cent. and at a sulphur content which increased with the amount of manganese present, there appeared a four-phase eutectic plane at which the separation of free FeS took place and of which the ternary eutectic point practically coincided with the binary eutectic of the Fe-FeS system.



FIGS. 4 TO 7—*Sections Across the Iron-Manganese-Sulphur Diagram.*

The paper contains many interesting photomicrographs, including some which illustrate the condition of manganese sulphide in alloys with 0·5 per cent. of sulphur and 2 per cent. of manganese, but no reference is made to lower sulphur or manganese contents. For the correct understanding of the equilibrium relations in this region of the diagram, modifications required by the δ-γ change in iron would have to be taken into account, but this aspect of the subject is not touched upon in the present paper.

REFERENCES.

[1] *Arch. Eisenhüttenwes.*, 1935–36, 9, 247.
[2] *Id.*, 1936–37, 10, 581.
[3] *Id.*, 1937–38, 11, 41.
[4] *Id.*, 1932–33, 6, 495.
[5] *Stahl und Eisen*, 1936, 56, 433.
[6] *Arch. Eisenhüttenwes.*, 1934–35, 8, 187.
[7] *Id.*, 1935–36, 9, 535; Carnegie Schol. Memoirs, 1935, 24, 103.
[8] Fe-Mn, Gayler, *Journ.*, Iron and Steel Inst., 1933, 128, 293; FeS-MnS., Shibata, *Sci. Rep., Tôhoku Imp. Univ.*, 1928, 7, 279; Fe-FeS, Löbe and Becker, *Zeit. anorg. Chem.*, 1912, 77, 301.

Books and Publications.

The Alloys of Iron and Chromium: Vol. I, Low-chromium Alloys. By A. B. KINZEL and WALTER CRAFTS. 8vo, pp. xv+535. New York and London: McGraw-Hill Publishing Company, Ltd. 1937. 36s.

IN the series of monographs published under the auspices of the Engineering Foundation of America, the alloys of iron and chromium are to be dealt with in two volumes of which the present is the first. It therefore contains, in addition to a review of the properties of the chromium steels and cast irons with less than 10 per cent. of chromium, much information which is applicable to the whole range of chromium steels, namely, a historical introduction and chapters on the constitution of the iron-chromium alloys, the iron-chromium-carbon system, and transformations in chromium steels. These are followed by an extremely well-balanced and informative chapter of fifteen pages on the "General Effects of Chromium in Iron and Steel," which, in the authors' metaphor, serves as the map of the world which, in any atlas, precedes the maps of the separate continents and countries. The remaining 375 pages contain the detailed information on every aspect of the properties of the low-chromium steels and cast irons, which will make this volume an indispensable reference book to all who have occasion to deal with these products.

The commercial use of chromium steels for structural purposes dates back further than is sometimes supposed. Dr. H. M. Howe, in his treatise on the "Metallurgy of Steel," published in 1890, threw some doubt on the first reported use of chromium steel in the Eads Bridge over the Mississippi River at St. Louis in 1874, but tests reported by E. E. Thum (1928) showed that the staves for the main arch members of the bridge actually contained from 0·54 to 0·68 per cent. of chromium. The use of chromium in Bessemer steel tires, reported by Arnold in 1888, is also interesting in view of its much more recent introduction into rail steels.

The chapters on the constitution of the chromium steels deserve special commendation. There is a useful account of the evidence that has accumulated for the existence of the sigma phase (FeCr), and a most excellent summary of the work which has been done on the iron-chromium-carbon system, in the final assessment of which the results of the researches of Tofaute and his co-workers (1936) receive full consideration and acceptance. In this part of the work the authors have secured the assistance of Mr. J. S. Marsh, of the Alloys of Iron Research Staff of the Engineering Foundation, and they, as well as Mr. Marsh, are to be congratulated on the selection and arrangement of the contents of these important chapters. They clearly show that the general features of the iron, iron-carbide, chromium-carbide, chromium diagram are now well understood and that the details are better known than those of many other ternary systems.

In the section on critical ranges a comparison of Fig. 54 with the data of Maurer and Nienhaus (Table XIV) indicates that the diagrams for the 0·5 and 1·0 per cent. carbon steels should be reversed (though, it may be added, they appear exactly as in the original paper). As they stand at present, they show a separate $Ar_{3\cdot2}$ point in the 1·0 per cent. instead of in the 0·5 per cent. carbon steels, and a fall instead of a rise, with increase of carbon, in the limit of chromium content of the γ phase.

The value of the chapters on the mechanical and other properties of the low-chromium steels is enhanced by the authors' summaries usually found in books of this series. These will, in general, command hearty agreement, but occasionally there are conclusions which may be questioned. For example, the statement that "perhaps the most important of all the effects of chromium on the mechanical properties is the tendency to eliminate mass effect" (page 226) seems to exaggerate its influence in this direction, since it is well known that the presence of nickel or molybdenum or both is necessary in order to get uniform hardening in large masses even when the chromium exceeds 3 per cent.

The responsibility of chromium for susceptibility to temper brittleness is perhaps underestimated on page 193, in saying that "it is no more harmful than other alloys, excepting molybdenum." Without implying that they are "harmful," it may be recognised that there is a good deal of evidence to show that among the metals chromium and manganese are the most effective elements in rendering steel susceptible to temper brittleness. In this connection the undesirable term "impact strength" occurs in places to denote energy absorbed in fracture, though "impact resistance" and "impact value" are also frequently employed in the book. The property measured cannot be regarded as "strength." The impact value is, in fact, far more dependent on a "ductility" factor than on a "strength" factor.

A large part of the book refers to the properties and uses of "modified" chromium steels, *i.e.*, steels containing chromium and an additional element, such as nickel, manganese, silicon, copper, vanadium, molybdenum, &c. Here and elsewhere the authors have been able to introduce hitherto unpublished results of original work carried out by themselves and their colleagues in the Union Carbide and Carbon Research Laboratories.

This latest addition to the Alloys of Iron Series fully maintains the high standard of the previous volumes, and will remain for a long time the standard work of reference on the low-chromium steels.